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Unusual through-thickness variations of microstructure and texture in heavily rolled and annealed Al–0.3%Cu

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Abstract

The evolution of microstructure and texture has been investigated in coarse-grained Al–0.3%Cu subjected to 98% cold rolling and subsequent annealing for 1 h at temperatures up to 300 °C. In the rolled material the distribution of crystallographic orientations is found to be highly non-uniform through the sample thickness, with the center layer being strongly dominated by the Bs component, whereas no single texture component dominates in the subsurface. The rolled microstructure contains lamellar structures and a large number of shear bands in all layers. The shear bands promote

nucleation of Goss-oriented and other grains in the center layer. Most grains nucleated in the subsurface have random orientations. In the fully recrystallized microstructure obtained after annealing at 225 °C for 1 h, a strong Goss texture is formed in the center layer, while random orientations constitute the greatest area fraction of all orientations in the subsurface. Grain growth taking place after complete recrystallization at 250 °C and 300 °C leads to unusually high fractions of the Goss texture and high fractions of low angle boundaries in the center layer. In the subsurface layers, grain growth results in increased fractions of random orientations, and consequently most grain boundaries in these layers are high angle boundaries. The sample annealed at 300 °C for 1 h exhibits a pronounced sandwich-type structure, with a Goss-oriented band in the center layer enclosed by subsurface layers containing coarser grains of many different orientations.

Keywords: aluminum alloys; cold rolling; annealing; Goss texture; microstructure heterogeneity; scanning electron microscopy

1. Introduction

Thermo-mechanical processing of metallic materials frequently results in sample-scale heterogeneities of the microstructure and crystallographic texture [1–4]. This is especially true for materials processed by rolling, where through-thickness gradients of the microstructure and texture are usually observed [5–14]. Both friction and roll gap geometry control the distribution of strain through the sample thickness, which can induce different through-thickness gradients in rolled materials. The most commonly reported sample-scale heterogeneity after conventional rolling is associated with an increased shear strain near the surface of rolled products. In face centered cubic materials, this strain results in a so-called shear texture dominated by the $\{001\}\langle 110 \rangle$ component,

whereas the center layer typically demonstrates strong texture components characteristic of plane strain compression, i.e. Brass (Bs) $\{110\}\langle 112\rangle$, S $\{123\}\langle 634\rangle$ and Copper (Cu) $\{112\}\langle 111\rangle$ components. Upon annealing such deformation-induced gradients inevitably lead to heterogeneities in the recrystallized microstructure and texture [13,14], which can affect the mechanical behavior of the product.

The development of texture gradients during conventional rolling as a function of roll gap geometry and friction coefficient has been described in a large number of publications, e.g. [5–11]. Other texture gradients can develop during asymmetric rolling [14–16] or due to heterogeneities present in the material before rolling, such as variations in the initial texture or/and initial grain size. In particular, it is known that the evolution of microstructure and texture during rolling of samples with a coarse initial grain size can be substantially different from those in initially fine-grained samples, leading also to significant differences in their annealing textures [17,18].

In several recent studies of a heavily cold-rolled Al–0.3%Cu alloy with a large initial grain size, an unusually strong Bs texture was observed in the center layer of the deformed material, while a very strong Goss texture developed in this layer during recrystallization and grain growth [19–21]. Since rather atypical microstructures and textures were observed in the center layer of this alloy, it is of interest to investigate whether this material also contains unusual through-thickness gradients. In the present work, an investigation of through-thickness variations of the microstructure and texture is conducted for the Al–0.3%Cu alloy cold-rolled to 98% and annealed isochronally at different temperatures.

2. Experimental

An Al–0.3%Cu alloy with 1.5–2 mm large grains in the as-cast condition was forged at 200 °C to obtain a 50 mm thick plate suitable for cold rolling. The texture of this forged sample contained an increased intensity of orientations near the Goss component [19]. This sample was 98% cold-rolled unidirectionally by 15 passes with no lubrication to a final thickness of ~1 mm (von Mises strain $\varepsilon_{VM} = 4.5$). The l/h ratio, where l is the chordal length of the contact between the rolls and the specimen, and h is the mean sample thickness [9], was maintained within the range 3 to 4. The sample was then annealed in air at temperatures in the range 100–300 °C.

For each condition, a section containing the rolling direction (RD) and the normal direction (ND) was polished mechanically and then electrochemically in a Struers A2 electrolyte at –15 °C. The polished sections were investigated using a Zeiss Supra 35 field emission gun scanning electron microscope (SEM) and a Zeiss AURIGA dual-beam SEM. Backscattered electron (BSE) images were taken to reveal second phase particles present in the microstructure. The microstructure and crystallographic texture were then investigated using electron backscatter diffraction (EBSD). EBSD data were collected at accelerating voltages of 15 kV and 20 kV, applying a step size of 25–30 nm for investigations of the deformed microstructure and a step size of 0.5–2 μm for studying grains in the partially and fully recrystallized conditions. For each sample, several orientation maps from different regions were collected to enable statistically sound data. Low angle boundaries (LABs) and high angle boundaries (HABs) were defined in these maps as those with misorientation angles $\theta = 2\text{--}15^\circ$ and $\theta \geq 15^\circ$, respectively. Recrystallized grains were defined as regions greater than 5 μm with internal misorientations less than 1° , at least partly surrounded by HABs [22].

Textures were studied through the sample thickness with a step size of either 4 μm or 5 μm . Each through-thickness data set was divided into 3 subsets, corresponding to three layers of equal thickness: one center layer and two opposite subsurface layers. The data obtained from the two subsurface layers were averaged during quantitative analysis. Fractions of different texture components were calculated applying a 15° deviation from the closest exact $\{hkl\}\langle uvw \rangle$ orientation. In addition to the fractions of the Goss, Q $\{013\}\langle 231 \rangle$, Bs, S and Cu components analyzed in our previous publication [20], fractions of the shear (Sh) $\{001\}\langle 110 \rangle$ and ND-rotated cube (CND) $\{001\}\langle 310 \rangle$ were also calculated in the present experiment, as the latter components became prominent in the recrystallized subsurface layers. Vickers hardness was measured along the mid-thickness in the RD-ND section with a load of 500 gf and a dwell time of 15 s.

3. Results

3.1 Cold-rolled material

Figure 1 presents orientation distribution functions (ODFs) calculated from the EBSD data in four different locations of the sample, covering a total area of $\sim 3 \text{ mm}^2$. It is obvious that the Bs component strongly dominates the crystallographic texture in the center layer (see Fig.1a). Analysis of the data indicates that in the center layer the area fraction of the Bs component (f_{Bs}) is 65%, whereas area fractions of the S, Cu and Goss (G) components are 14%, 5% and 6%, respectively (see Table 1). A very different rolling texture is observed in the subsurface (Fig.1b). Here the fractions of the Bs, S and Cu components are approximately 20–30% each, while the fraction of the G texture is only 3%. The fraction of the $\{001\}\langle 110 \rangle$ shear texture component in the subsurface is $<1\%$. The $\{001\}\langle 110 \rangle$ component is similarly weak very near the sample surface, as determined from the EBSD data

collected in the rolling plane after only gentle electropolishing.

Examination of several regions in the rolled material reveals considerable local variations in the microstructure, which reflects the extremely large initial grain size. Example orientation maps in Fig.2 capture the key differences between the different layers of the deformed sample, namely the predominance of the Bs component in the center and a more uniform distribution of the three major rolling texture components in the subsurface. It is seen that each layer contains lamellar structures intersected by shear bands at approximately 20° to 40° to the RD (see Fig.2 and Fig.3). The average boundary spacing along the ND (d_{ND}) is slightly smaller in the center, 212 nm, than in the subsurface, 242 nm (Table 1). Fine Al_2Cu particles (seen as bright features in BSE images) are present along some boundaries in the microstructure, see Fig.3. The frequency of such particles is, however, rather small in all layers of the as-rolled sample.

Table 1 also presents average fractions of HABs (f_{HAB}) in the different layers. It is seen that in the strongly textured center layer f_{HAB} is rather small, 31%. In the subsurface layer, the fraction of HABs is significantly larger, 62%.

Table 1. Fractions of different texture components and parameters of the microstructure as measured using EBSD in the rolled material

Layer	f_{Bs} (%)	f_s (%)	f_{Cu} (%)	f_G (%)	d_{ND} (nm)	f_{HAB} (%)
Center	65	14	5	6	212	31
Subsurface	27	32	22	3	242	62

3.2 Annealed samples

To select samples for microstructure and texture examinations, Vickers hardness was measured in a large number of samples annealed for 1 h in the temperature range 100–300 °C. The results of these measurements are shown in Fig.4 along with the hardness of the as-rolled material. It is evident that the hardness decreases from 66 HV in the as-rolled condition to 46 HV after annealing at 150 °C and that the slope of the softening curve changes significantly between 150 °C and 200 °C, followed by an almost constant hardness of 21–23 HV between 200 °C and 300 °C. The shape of this curve suggests that the softening at temperatures up to 150°C is mostly due to recovery and that recrystallization takes place at temperatures above 150 °C. The same conclusion was also made by Huang et al. [21], who studied the influence of structural parameters in the center of the same sample on strength. Based on the softening curve, six samples were chosen for SEM characterizations.

Figure 5 shows the microstructure in the center of the sample annealed at 150 °C along with the microstructure of the as-rolled sample to facilitate comparison between these two conditions. It is obvious that the annealed microstructure is significantly coarser and contains a much higher frequency of Al₂Cu particles than that in the rolled sample (cf. Fig.5a and Fig.5b). The particles precipitate along the lamellar boundaries, so that previous boundary locations decorated by the precipitates can be easily traced in the coarsened microstructure. The average boundary spacing along the ND is approximately 450 nm in the center layer and 480 nm in the subsurface. Despite the two-fold increase in d_{ND} as compared to that in the as-rolled sample (see Table 1), the lamellar morphology is preserved in this annealed condition. No recrystallized grains were found in the areas examined.

Coarsening of the deformed microstructure and precipitation also take place during annealing at 175 °C. However, the microstructure annealed at this temperature already contains recrystallized grains (see Fig.6 and Fig.7). While the average size D_{Rex} of such grains is similar ($\sim 10 \mu\text{m}$) in the center and subsurface layers, the fraction of the recrystallized microstructure f_{Rex} is slightly different in these layers, 3% in the center and 8% in the subsurface (averaged over the two opposite layers). Recrystallization is seen to initiate in shear bands, where grains of Goss, Q and random orientations are found more frequently than grains of other orientations, and within horizontal bands with a large variety of orientations. Furthermore, it is found that very small recrystallized grains are typically particle-free, whereas comparatively large grains always contain arrays of particles aligned with lamellar boundaries that were present in the deformed microstructure before annealing (see Fig.7). This observation suggests that during annealing at 175 °C nuclei first appear in particle-free regions and that further recrystallization occurs concurrently with precipitation.

After 1 h at 200 °C, f_{Rex} is already 67% and 71% in the center and subsurface layers, respectively (see Fig.8a and Fig.9a) and D_{Rex} is 14–15 μm (Fig.8b). The pronounced recrystallization results in significant changes in texture. The Bs orientation is no longer the strongest texture component in the center layer, and fractions of the Goss, Q and random orientations are all higher than those in the as-deformed condition and after annealing at 175 °C (see Fig.10a). In the subsurface, the Bs, S and Cu components weaken during annealing at 200 °C, while fractions of the other texture components increase (Fig.10b).

At temperatures in the range 225–300 °C the microstructure is fully recrystallized (see Fig.8a and Fig.9b,c). The increase in the annealing temperature from 225 °C or 250 °C results in mostly

quantitative changes in the microstructure, such as an increase in the size of recrystallized grains in the sample annealed at 250 °C (Fig.9b) and a greater area fraction of the Goss texture in the center layer, where it reaches 64% (Fig.10a). The difference between the samples annealed at either 250 °C or 300 °C is more remarkable. Firstly, the average grain size D_{ReX} in the subsurface of the sample annealed at 300 °C is increased by a factor of almost 3 as compared to that in the sample annealed at 250 °C (Fig.8b). Whereas the average D_{ReX} in the subsurface of the sample annealed at 300 °C is ~60 μm , a number of subsurface grains in this sample even reach a size of 300 μm . Such large grains are not observed after annealing at 250 °C (cf. Fig.9b and Fig.9c). Secondly, there are significant differences in the spatial distribution of different texture components between these two samples. In contrast to the sample annealed at 250 °C, where central Goss-oriented grains are frequently interspersed with grains of other orientations, there are very few non-Goss grains near the mid-thickness of the sample annealed at 300 °C. Moreover, subsurface grains penetrate into the center layer of the latter sample, so that the fraction of the Goss texture here is even somewhat lower than that after annealing at 250 °C (see Fig.10a). In fact, in the sample annealed at 300 °C, the Goss component is confined to a 200–300 μm wide band in the center layer. The subsurface grains in this sample are characterized by a higher fraction of random orientations and a lower fraction of the Goss-texture than those in the sample annealed at 250 °C. The area fractions of the shear, Q and CND components in the subsurface of the 300 °C sample are also significant, 12% to 17% each. Thus, the sample annealed at 300 °C exhibits a very pronounced sandwich-type structure, with the Goss-dominated central band and two subsurface layers containing grains of mostly random, shear, Q and CND orientations (Fig.9c and Fig.10). The grain boundary populations are also very different in the

different layers of this sample. The majority of all boundaries in the center layer are LABs, whereas HABs dominate in the subsurface (Fig.9c).

4. Discussion

The material studied in the present work demonstrates rather unusual through-thickness variations in the cold-rolled and annealed conditions, clearly different from those described in Refs.[5–14]. The center layer of our deformed material is characterized by a very high fraction of the Bs component, which is replaced by an exceptionally strong Goss texture in the samples annealed at 250 °C and 300 °C for 1 h. On the other hand, the deformation texture in the subsurface contains well-developed Bs, S and Cu components and no significant shear texture. In these subsurface layers, high fractions of random orientations are found after recrystallization and grain growth. The development of these unusual through-thickness heterogeneities during rolling and then annealing is considered in the following.

4.1 Through-thickness heterogeneity of the rolled material

In our previous paper [20], the dominance of the Bs component in the center layer of a 95% rolled Al–0.3%Cu sample has been attributed to combined effects of the initial texture (increased intensities of near-Goss orientations) and the large initial grain size. The initial texture is known to have a significant effect on the final rolling texture. In particular, as shown for pure aluminum [23], an initial Goss texture promotes a strong Bs texture during heavy rolling. The large initial grain size also makes the material prone to shear banding [17,18], which can further enhance the Goss texture with a subsequent rotation towards the Bs orientation during continued rolling to high strains [24]. In the extreme case, a very large grain size may even result in plastic deformation by predominantly single

slip, at least up to moderate strains. In such a case, for $\sigma_{RD} = -\sigma_{ND}$ and lattice rotations computed according to [25], only the exact Goss orientation is stable, while all grain orientations, including those slightly deviating from the exact Goss orientation, will rotate towards Bs [26]. In general, a starting texture composed of 80% random orientations and 20% near-Goss orientations would evolve by single slip into a texture with $f_{Bs} = 92\%$ and $f_G = 6\%$ after 98% rolling. It should of course be noted that single slip conditions induce significant strain incompatibilities not accounted for in the simulations and that flattening of the grains at high rolling reductions implies relaxation of only some shear components. The real deformation conditions are obviously more complex due to shear banding taking place during heavy rolling. Therefore, in a real texture of any coarse-grained material the final texture is expected to be significantly weaker than the one simulated based only on single slip. It is thus possible that a number of factors, i.e. the combination of the specific initial texture and very coarse grains promoting single slip and intense shear banding, all contribute to the dominant Bs texture observed in the center layer. They all agree with the texture change between the 95% rolled sample and the present sample rolled to 98%, where the fraction of the Bs component in the center layer increases from 51% to 65%, while the fraction of the Goss component decreases from 11% to 6% (see Ref.[20] and Table 1).

The different texture in the subsurface layer can be a result of a high frictional strain imposed by rolling without lubrication, i.e. RD-ND shear. This shear normally induces a strong $\{001\}\langle 110 \rangle$ texture component, which is, however, not observed in the present material. In line with the single slip simulations for the center layer, analogous simulations with addition of a shear component of 0.4 of the rolling strain predict a dominant S component with $f_S = 70\%$. This result is qualitatively

consistent with the observed difference between the two layers, again suggesting that the role of the large initial grain size may indeed be important.

The significant differences in the rolling texture govern the spatial distribution of orientations and hence boundary populations in the different layers, where the fraction of HABs in the subsurface is higher due to the more balanced proportion of the Bs, S and Cu components (see Table 1). These HABs are likely to migrate during dynamic recovery, which can explain the larger d_{ND} spacing seen in the subsurface as compared to that in the center (Table 1).

4.2 Nucleation of recrystallized grains

Recrystallization in the 98% rolled sample initially takes place at shear bands and in the horizontal bands with different orientations, as was also the case with the 95% rolled sample of the same material [20]. Shear bands are known to be preferential nucleation sites and the observed nucleation in these regions is not surprising. However, whereas nucleation at shear bands is usually considered to result in weak annealing textures [17,18], the texture formed in the center layer of Al–0.3%Cu is rather strong even in the sample recrystallized at 225 °C without appreciable grain growth (Fig.10a).

It has been previously shown that the Goss-oriented regions in heavily rolled Al–0.3%Cu are characterized by a lower stored energy than regions of the Bs texture, which creates a driving force for Goss-oriented subgrains to grow into the matrix of the dominant Bs component [20]. The reason for the different stored energies in regions of the different texture components can be understood based on the analysis of their Taylor factors M . These factors are a measure of the plastic work, and can thus be related to the stored energy [27,28].

For heavily rolled microstructures, Taylor factors are usually calculated assuming plane strain compression with relaxed shear components and neglecting shear banding. Table 2 shows M for several texture components in Al–0.3%Cu calculated for plane strain compression with relaxed RD-ND and TD-ND shear components, where TD is the transverse direction. The Bs, S and Cu components, all of which vanish during annealing, are characterized by high Taylor factors ($M = 3.3$ – 3.7) and hence presumably have higher stored energies. In contrast, Taylor factors for each of the components that become significant in the recrystallization texture are all considerably lower ($M < 3$) and so is the stored energy within these regions. It appears that the strength of the considered components in the annealing texture of the Al–0.3%Cu sample is governed by the differences in their Taylor factors and energy stored during deformation.

Table 2. Taylor factors M for several components identified in the rolling and annealing textures of Al–0.3%Cu

Bs	S	Cu	G	Q	Sh	CND
$\{110\}\langle 112 \rangle$	$\{123\}\langle 634 \rangle$	$\{112\}\langle 111 \rangle$	$\{110\}\langle 001 \rangle$	$\{013\}\langle 231 \rangle$	$\{001\}\langle 110 \rangle$	$\{001\}\langle 310 \rangle$
3.27	3.39	3.67	2.45	2.71	2.45	2.45

In general, grains nucleated prior to precipitation grow to a large size, presumably as these grains already reach a size large enough to overcome pinning before precipitation begins (see Fig.7). In the center layer, the Goss-oriented nuclei have a clear preference because they can nucleate both at shear bands, as initially proposed by Engler et al. [29], and within Bs/G bands where regions of the dominant Bs component are combined with Goss-oriented lamellae [20,30]. Though nucleation in

shear bands also takes place in the subsurface layers, many more recrystallized grains here appear to nucleate in the horizontal bands with a large number of random and other orientations (see Fig.6). As a result, a weak recrystallization texture is produced in these layers (see Fig.10b).

4.3 Evolution in the fully recrystallized samples

Comparing the fully recrystallized conditions, it is evident that the only sample, where the microstructure and texture are not significantly affected by grain growth, is the sample annealed at 225 °C for 1 h. In contrast to this sample, microstructures annealed at 250 °C and 300 °C appear to have experienced substantial grain growth after complete recrystallization. It can therefore be concluded that the unusually strong Goss texture observed in the center of the latter samples is obtained when Goss-oriented grains grow at the expense of other recrystallized grains. During this process, growing Goss-oriented grains form low energy LABs with grains of similar orientations, which enables a more efficient reduction in the overall grain boundary energy in this layer (see Fig.9). A similar texture strengthening during grain growth does not occur in the subsurface layers since there is no single dominant orientation in their recrystallization texture. In these layers, grain growth primarily results in increased fractions of random orientations (Fig.10b).

The changes in texture and in the spatial distribution of grains with different orientations greatly affect the proportion of LABs and HABs in the recrystallized conditions. Texture strengthening during grain growth typically results in increased fractions of LABs (f_{LAB}) [31], as is also seen for the center layer of the samples annealed at either 225 °C or 250 °C. In these two samples, f_{LAB} increases from 26% to 36% (Fig.11) as the Goss texture in the center strengthens from $f_{\text{G}} = 39\%$ to $f_{\text{G}} = 64\%$. Interestingly, f_{LAB} in the center layer increases even more, up to 66% (see Fig.11), as the annealing

temperature changes from 250 °C to 300 °C though there is no simultaneous increase in the fraction of the Goss texture. Obviously, the large difference in f_{LAB} in the center layer of the two samples annealed at either 250 °C or 300 °C reflect differences in the spatial distribution of orientations. In the center of the sample annealed at 250 °C, the Goss grains are interspersed with grains of other orientations, which results in many HABs. Therefore, f_{LAB} is only 36% in this sample. In the sample annealed at 300 °C, the spatial distribution of orientations is different. In the center of this sample, the Goss-oriented grains form one extended Goss band, where the vast majority of boundaries are LABs. HABs in this center layer are mostly observed on the periphery of the central Goss band, so that their fraction is significantly smaller than f_{LAB} .

Considering the subsurface layers, there is only a minor increase in the LAB fraction with increasing annealing temperature, and f_{LAB} remains comparatively low, 22%, even at 300 °C as a result of many different grain orientations present in these layers. During annealing at 300 °C, the reduction in the grain boundary energy is achieved in these layers by grain growth both within the subsurface regions and into the center layer. Taking into account that the average size of the subsurface grains is larger than that of the central grains (see Fig.8b), it is reasonable to expect that if annealing at 300 °C is continued beyond 1 h, the Goss-dominated layer in the center would largely be consumed by grains growing from the opposite subsurface layers.

5 Conclusions

1. Heavy rolling of an initially coarse-grained Al–0.3%Cu sample results in a highly non-uniform distribution of crystallographic orientations through the sample thickness. The center layer is strongly dominated by the Bs component, while a more balanced proportion of the Bs, S and

Cu components is found in the subsurface layers. These differences in the deformation texture lead to very different boundary populations in the different layers, where the fraction of HABs after 98% rolling is 31% in the center and 62% in the subsurface. Each layer contains lamellar structures intersected by a large number of shear bands. This type of microstructure and texture are retained at temperatures below 175 °C when only recovery accompanied by structural coarsening and precipitation take place.

2. Recrystallization occurs during annealing for 1 h at a temperature of 175 °C and above. As is evident from microstructural observations at an early stage of recrystallization (in the sample annealed at 175 °C), first recrystallized grains nucleate at shear bands and within horizontal bands with a large variety of orientations. The shear bands promote nucleation of Goss and other grains in the center layer. Nucleation in the subsurface predominantly generates grains of many different orientations.
3. The microstructure is fully recrystallized after annealing for 1 h at temperatures ≥ 225 °C. As the annealing temperature changes from 225 °C to 250 °C, the fraction of the Goss texture in the center increases from 39% up to 64% and the fraction of LABs increases from 26% to 36% as a result of grain growth taking place after complete recrystallization. In the subsurface layers of each recrystallized sample, random orientations are more frequent than orientations of any other component, and the majority of all grain boundaries are HABs.
4. The sample annealed at 300 °C exhibits a pronounced sandwich-type structure, with a continuous Goss-oriented band in the center layer and two subsurface layers containing grains of other orientations. The fraction of LABs is 66% in the center and 22% in the subsurface

layers. The subsurface grains are coarser than those in the center and demonstrate the tendency to penetrate into the center layer.

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Figure captions

Figure 1. ODF sections ($\phi_2=0^\circ$, 45° and 65°) representing texture in the center (a) and subsurface (b) layers of the 98% rolled Al–0.3%Cu sample. Contour levels are 1, 3, 5, 10, 20, 30, $40 \times$ random. Positions of the Bs, S, Cu and G texture components are shown in (c).

Figure 2. Example orientation maps showing the microstructure in the center (a) and subsurface (b) of the 98% rolled Al–0.3%Cu sample. Orientations that do not belong to any orientation listed in the color code are represented by different shadings of gray. LABs and HABs are shown as white and black lines, respectively. The RD is parallel to the scale bar. (For interpretation of the references to color in this figure legend, the reader is referred to the Web version of this article.)

Figure 3. BSE image from the center layer of the 98% rolled Al–0.3%Cu sample showing a shear band intersecting lamellar structures. Small bright features along some lamellar boundaries are Al_2Cu particles.

Figure 4. Evolution of Vickers hardness during annealing of the 98% rolled Al–0.3%Cu sample for 1 h at different temperatures.

Figure 5. Effect of low-temperature annealing on the microstructure of the 98% rolled Al–0.3%Cu sample: (a) as-rolled condition; (b) after annealing at 150°C for 1 h.

Figure 6. Orientation maps showing recrystallized grains in the sample annealed at 175°C for 1 h: (a) center; (b) subsurface. Orientations that do not belong to any orientation listed in the color code are represented by different shadings of gray. LABs and HABs are shown as white and black lines, respectively. The RD is parallel to the scale bar. (For interpretation of the references to color in this

figure legend, the reader is referred to the Web version of this article.)

Figure 7. BSE images showing precipitates in partially recrystallized regions of the sample annealed at 175 °C for 1 h: (a) center; (b) subsurface.

Figure 8. Changes in the area fraction (a) and average size (b) of recrystallized grains during annealing at different temperatures for 1 h.

Figure 9. Orientation maps covering almost the entire thickness of the Al–0.3%Cu sample annealed for 1 h at different temperatures: (a) 200 °C (a); (b) 250 °C and (c) 300 °C. Orientations that do not belong to any orientation listed in the color code are represented by different shadings of gray. LABs and HABs are shown as white and black lines, respectively. The RD is parallel to the scale bar. (For interpretation of the references to color in this figure legend, the reader is referred to the Web version of this article.)

Figure 10. Fractions of different texture components in the as-rolled and annealed conditions: (a) center; (b) subsurface.

Figure 11. Fractions of LABs in the fully recrystallized samples.

G Q Sh CND Bs S Cu

10 μm

(a)

(b)

G Q Sh CND Bs S Cu

50 μm

(a)

(b)

G Q Sh CND Bs S Cu